09/355987

PCT/AU98/00115

AUSTRALIA

REC'D 17 MAR 1998
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KIM MARSHALL
MANAGER EXAMINATION SUPPORT AND SALES

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AUSTRALIAN
PROVISIONAL No. DATE OF FILING
P05268 24 FEB. 97
PATENT OFFICE

PROVISIONAL SPECIFICATION

FOR THE INVENTION ENTITLED:

"IMPROVED FOUNDRY ALLOY"

Applicant:

CAST CENTRE PTY LTD

The invention is described in the following statement:

IMPROVED FOUNDRY ALLOY

The present invention relates to an improved foundry alloy and to a method of producing an improved foundry alloy. In particular, the improved foundry alloy is an aluminium-based alloy.

Primary metal based foundry alloys are largely used for structural or safety type applications where there is a requirement for high and consistent mechanical properties. The majority of components made from aluminium foundry alloys are made from hypoeutectic aluminium-silicon-magnesium alloys containing a nominal silicon level of 7% by weight (601 and 603 designations). In simple terms these alloys are a composite of hard, discontinuous silicon particles and large, brittle iron intermetallics embedded in a ductile aluminium matrix.

There are three registered Australian compositions for strontium modified aluminium-7% silicon alloys. These are given in Table 1. The magnesium content of the alloys covers the range 0.25 to 0.4 (601 alloys) and 0.45 to 0.7 (603 alloys) weight percent. The addition of magnesium allows castings to be heat treated to form magnesium silicide precipitates. These harden the matrix of the alloy to obtain the desired combination of strength and ductility.

Table 1: Registered alloy composition for strontium modified 601/603 type foundry alloys

Alloy Code	Si	Fe	Cu	Mn	Mg	Zn	Ti	Other Each	Other Total	Al
AC601	6.5-7.5	0.20	0.05	0.05	0.30-0.40	0.05	0.20	0.05	0.15	Rem
CC601	6.5-7.5	0.20	0.05	0.05	0.25-0.35	0.05	0.20	0.05	0.15	Rem
AC603	6.5-7.5	0.15	0.05	0.03	0.45-0.7	0.05	0.20	0.05	0.15	Rem

♦ Compositions in weight percent. Compositions indicate a maximum unless a range is given

The main impurity found in these alloys is iron. The iron solidifies into a number of brittle phases from the eutectic liquid. The two major iron bearing phases found in these alloys are the π phase ($Al_8Si_6Mg_3Fe$) which is the predominant phase formed in high Mg content alloys and the β phase (Al_5SiFe) which forms in low magnesium content alloys. The π phase forms into a script morphology while the

β phase is less voluminous but forms into acicular plates. Both phases are detrimental to mechanical properties. High Mg contents are desirable to provide higher strength, but the presence of π phase at high Mg contents causes the ductility of the alloy to unfavourably decrease.

Conventional theories on the micro-mechanics of failure of premodified 601 and 603 alloys state that the iron rich intermetallic phases are critical in determining the fracture toughness as the silicon particles are small and round. Increases in the magnesium content of these alloys increase the amount of the π phase, which has a negative impact on the ductile properties of the alloys. Further, as some magnesium is contained in the π phase, the maximum volume fraction of magnesium silicide precipitates cannot be obtained. Thus, the alloys do not achieve the maximum possible strength consistent with their magnesium content. Also, as the magnesium content of an alloy increases the magnesium content of the π phase may change leading to even greater volume fractions of the phase for a given Fe content. It is thus concluded that the overall quality of an alloy, as given by the quality index, decreases as increasing volume fractions of the π phase forms at increased magnesium contents. The quality index is given by the formula:

$$Q.I. = UTS + 150 \ell og_{10} E$$

where:

Q.I. = Quality Index (MPa)

UTS = Ultimate Tensile Strength (MPa)

E = Elongation at Fracture (%)

Attempts have been made to eliminate the π phase and thus remove its detrimental impact on mechanical properties. Beryllium additions can be used to precipitate the iron impurity as part of a $BeSiFe_2Al_8$ phase. This beryllium containing phase forms in preference to the π phase, leading to alloys with improved mechanical properties. Unfortunately, there are serious health hazards associated with using beryllium. Consequently, beryllium modification is not widely practised and the deleterious effect of the π phase on alloy quality remains. Other attempts to modify the Fe bearing phases by using say Mn have been tried in secondary

alloys but have not been used in primary alloys.

It is an object of the present invention to provide an improved foundry alloy.

In a first aspect, the present invention provides an alloy comprising:

Si	:	6.5 - 7.5 wt%
Fe	:	up to 0.20 wt%
Cu	:	up to 0.05 wt%
Mn	:	up to 0.05 wt%
Mg	:	0.35 to 0.50wt%
Zn	:	up to 0.05wt%
Ti	:	up to 0.20wt%

Balance : Al and other components, the other components comprising a total of not more than 0.15wt% and any single component of the other components not exceeding 0.05wt%, the alloy having a microstructure including a primary aluminium - containing matrix having dendritic arm spacing of about 10-45μm and iron-containing phases, wherein the iron containing phases comprise β phase and π phase.

Preferably, the iron-containing phases comprise β phase and a small amount of π phase. Preferably, the iron-containing phases include π phase in an amount of 10 vol% to 30 vol% of the iron containing phases, with the balance substantially being β phase, although the amount of π phase may be higher if the Mg content is in the upper end of the range.

The Mg content of the alloy is preferably about 0.40-0.45 wt%.

In the present invention, it has been realised that close control of the magnesium content of aluminium-silicon foundry alloys, in such a way that only a small amount of π phase forms, can actually lead to an increase in alloy quality. This increased alloy quality corresponds to an alloy with improved mechanical properties. Further, when the magnesium content is controlled to obtain the maximum quality, the variation in alloy quality for a small change in magnesium level is minimal. Thus, the consistency in the mechanical properties of the alloy is maximised. Therefore, the new invention provides a variant of the 601/603 type

foundry alloys which has improved and more consistent mechanical properties, as determined by alloy quality.

The present invention also provides a method for manufacturing an alloy article.

In a second aspect, the present invention provides a method for manufacturing an alloy article comprising:

- providing a melt having a composition of:

Si	:	6.5 - 7.5 wt%
Fe	:	up to 0.20 wt%
Cu	:	up to 0.05 wt%
Mn	:	up to 0.05 wt%
Mg	:	0.35 to 0.50wt%
Zn	:	up to 0.05wt%
Ti	:	up to 0.20wt%

Balance : Al and other components, the other components comprising a total of not more than 0.15wt% and any single component of the other components not exceeding 0.05wt%,

- casting said melt and solidifying the casting at a cooling rate that achieves a dendritic arm spacing of about 10-45μm,
- solution heat treating the alloy and then quenching the alloy,
- wherein the alloy article has a microstructure that includes iron-containing phases comprising β phase and π phase.

Preferably, the iron containing phases comprise β phase and a small amount of π phase. Preferably, the iron-containing phases in the alloy article include π phase in an amount of 10 to 30 vol%, with the balance of the iron-containing phases substantially comprising β phase. However, higher levels of π phase may be present if the Mg content is at the upper end of the above range.

It is preferred that the cooling of the casting produces iron-containing phases in the alloy microstructure that includes a substantial proportion of π phase and the subsequent solution heat treatment is effective to convert at least some and

preferably a majority of the π phase to β phase to give a microstructure in the alloy article including iron-containing phases comprising β phase and a small amount of π phase.

The melt prior to casting may be at a temperature above the liquidus temperature of the alloy, with the melt having sufficient superheat to fill the mould, that is at a temperature of 680-720°C.

The solution heat treatment preferably involves heating the casting to about 540°C and holding the casting at that temperature for about 8 hours.

After the solution heat treatment, the casting is preferably quenched, more preferably quenched in hot water, such as hot water having a temperature of 70-80°C. After quenching, the alloy article is cooled to room temperature and optionally subjected to an aging heat treatment. The aging heat treatment may involve heating the article to a temperature of 140-170°C and holding at that temperature for about 1-10 hours. After the aging heat treatment, the article may be air cooled to room temperature.

It is noted that the solution treatment of the alloys also has some effect on the composition at which a small amount of π phase is present in the alloy. This is because some of the π phase can transform to the β phase during solution treatment. This occurs less at higher magnesium contents which have larger volume fractions of π phase. Longer solution treatment times and higher solution treatment temperatures are preferred, with good results being achieved by solution heat treating at 540°C for 2 to 8 hours.

Results to support the present invention are given in Figure 1, where plots of typical response surfaces derived from experimentally determined quality index data are shown. The three surfaces correspond to alloys cast at different solidification rates. Solidification rate is commonly measured by the as-cast dendrite cell size or secondary dendrite arm spacing (DAS) but other methods exist. The results here use secondary dendrite arm spacing to indicate solidification rate, with a small dendrite arm spacing corresponding to a high solidification rate.

It can be seen that at high solidification rates ($\approx 20\mu$m DAS), the alloy quality

peaks at a magnesium level of 0.45-0.50 weight percent, at the intermediate solidification rates (≈40μm DAS) the quality peaks at a magnesium level of 0.35-0.40 weight percent, while at the low solidification rates (≈60μm DAS) the quality maximum occurs at a magnesium level of 0.25-0.30 weight percent. Further, the magnesium level for the peak quality is independent of iron, for the iron levels examined. Also, the curvature of the response surfaces is least near the peak in quality index.

The peak quality from Figure 1 seems to correspond well with microstructural evidence for the formation of a small amount of π phase. By increasing the magnesium content of the alloy, it can be seen that in some circumstances improved quality results. Further, under these conditions of improved quality, the response surfaces show less curvature and therefore the mechanical properties of the alloy are more consistent.

It should be noted that the present invention works best with those casting designs or casting methods which produce high solidification rates (≤45μm DAS) such as permanent mould, mould chill methods with sand and squeeze casting. Indeed, the trend in the automotive industry is to move away from thick section, low solidification rate (high DAS) castings towards smaller, lightweight castings with thinner sections and higher solidification rates (low DAS). The common belief prior to the present invention was that low magnesium levels produce higher quality castings. The results shown here confirm this to be true at low solidification rates (Figure 1c). However, at higher solidification rates, the magnesium contents covered by this invention show, surprisingly, improved alloy quality and therefore improved mechanical properties.

Figures 2(a) to 2(d) show photomicrographs of alloys. In Figure 2(a), the Mg content of the alloy is higher than the Mg content of the alloy of the present invention. The main phases shown in Figure 2(a) are spheroidal silicon-containing phases and the iron-containing π phase. Figure 2(b) shows the microstructure of an alloy containing less Mg than the alloy of the present invention. The phases present include spheroidal silicon-containing phase and iron-containing β phase. The β

phase is present as rod-like structures dispersed throughout the matrix. Figure 2(c) shows the microstructure of an alloy of the present invention. The phases include spheroidal silicon-containing phases, a small amount of π phase and β phase. The β phase is present as rod-like structures clumped together. This is consistent with the β phase being formed from transformation of π phase during heat treatment. Figure 2(d) shows the microstructure of the as-cast alloy of Figure 2(c), i.e. before heat treatment. This photomicrograph shows regions of π phase that are largely transformed to β phase during heat treatment.

The drive for alloys with improved mechanical properties stems from the major restraint that mechanical properties place on the design of a casting, or even if a cast alloy can be used to manufacture a certain component. The thickness of critical sections needs to be sufficiently large that the cast component can operate without failure. Mechanical properties of alloys therefore limit the minimum weight of a cast component. Further, the thickness of a casting's sections will determine the time required for the casting to solidify. For certain casting methods, such as low pressure die casting, the production rate is often determined by the casting's solidification rate as the casting machine is tied up until the casting has fully solidified. Finally, the solution treatment, quench rate and ageing treatment of a cast component may be tailored to its design so as not to induce unnecessarily high residual stresses. High residual stresses can cause distortion of the component requiring additional machining. The mechanical properties of the base alloy therefore affect all stages of a casting's manufacturing from design, to casting the component, the heat treatment, machining, final weight and production rate.

The present invention therefore has the more specific applications:

New markets for aluminium-7% silicon foundry alloys. Cast alloys generally have inferior mechanical properties but lower manufacturing costs to similar components made from wrought alloys. The high mechanical property requirements of some components necessitates the use of wrought alloys. This achievement of higher and more consistent mechanical properties may allow the use of this 601/603 alloy variant to replace wrought alloys, or other cast alloys, for some components.

Cast components with thinner sections and lower total weight. The improved and more consistent mechanical properties of the alloy of this invention allows components with thinner sections to be designed and cast. These components can still operate without failure and will have a lower total weight.

Cast components with an improved production rate. Castings with thinner sections may require less time to solidify. Production equipment will then be tied up for less time waiting for a component to solidify. The production rate is thus increased.

Cast components with refined iron and silicon intermetallic phases. The solidification time of a casting strongly determines the coarseness of the microstructure. Components with thinner sections and therefore higher solidification rates (and lower solidification times) will have a more refined microstructure. This refining of the microstructure will provide additional improvements to the mechanical properties of a casting, independent of the use of a superior alloy.

Cast components with reduced heat treatment time. Castings with thinner sections require less time to homogenise. Further, the time required for the casting to reach the solution treatment temperature or ageing temperature will be less. This also benefits the production rate of components.

Cast components with increased quench rate. Thinner castings may quench more rapidly. This may lead to improved mechanical properties as it suppresses the formation of magnesium-silicide particles during cooling. These improved properties are independent of any refinement of the microstructure or the use of a superior alloy.

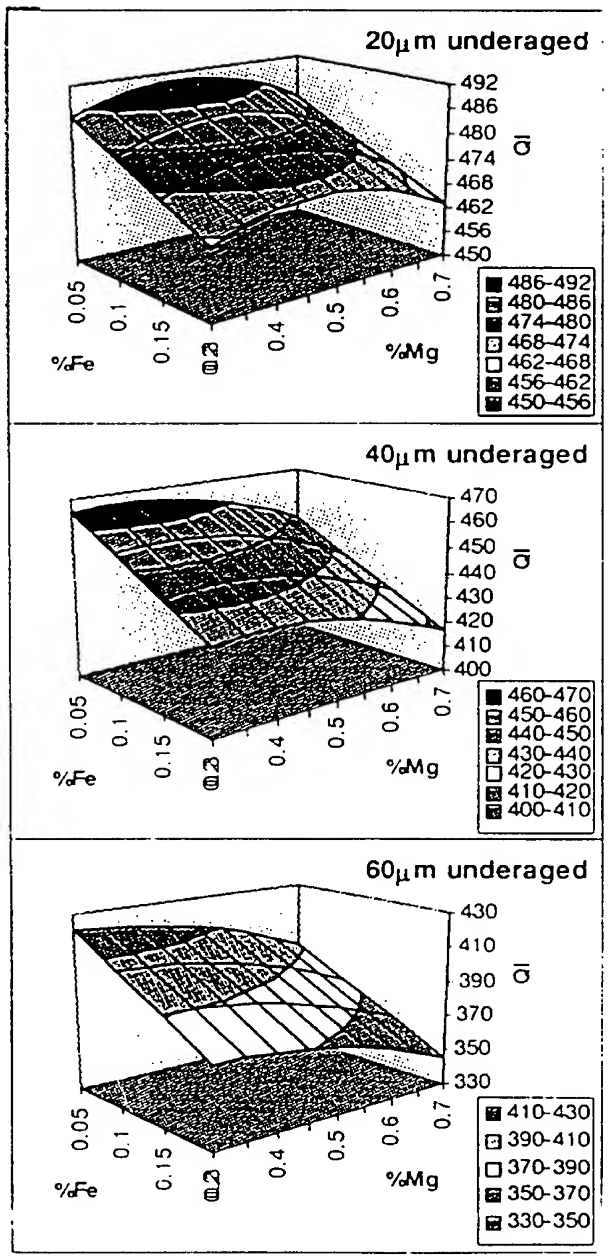
It will be appreciated that the invention described herein is susceptible to variation and modifications other than those specifically described. It is to be understood that the invention encompasses all such variations and modifications that fall within its spirit and scope.

DATED: 24 February 1997

CARTER SMITH & BEADLE

Patent Attorneys for the Applicant:

CAST CENTRE PTY LTD



20μm underaged
492
486
480
474
468
462
456
450
QI
0.05
0.1
0.15
0.3
0.4
0.5
0.6
0.7
%Fe
%Mg
486-492
480-486
474-480
468-474
462-468
456-462
450-456
40μm underaged
470
460
450
440
430
420
410
400
QI
0.05
0.1
0.15
0.3
0.4
0.5
0.6
0.7
%Fe
%Mg
460-470
450-460
440-450
430-440
420-430
410-420
400-410
60μm underaged
430
410
390
370
350
330
QI
0.05
0.1
0.15
0.3
0.4
0.5
0.6
0.7
%Fe
%Mg
410-430
390-410
370-390
350-370
330-350

Figure 1

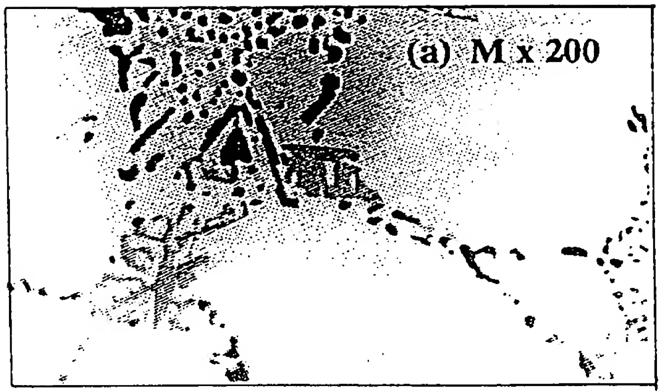


Figure 2a

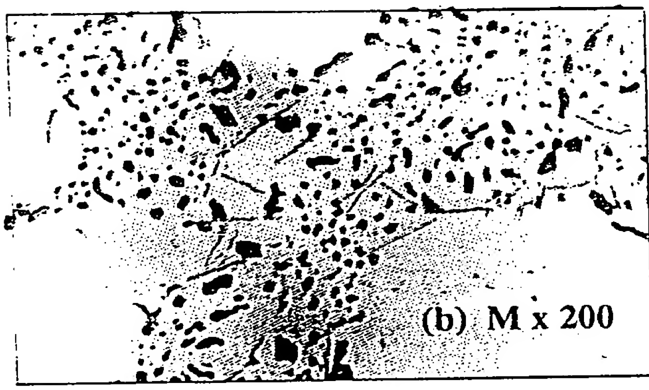


Figure 2b

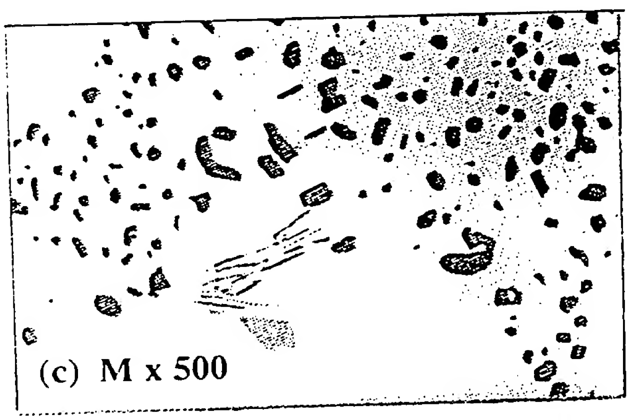


Figure 2c

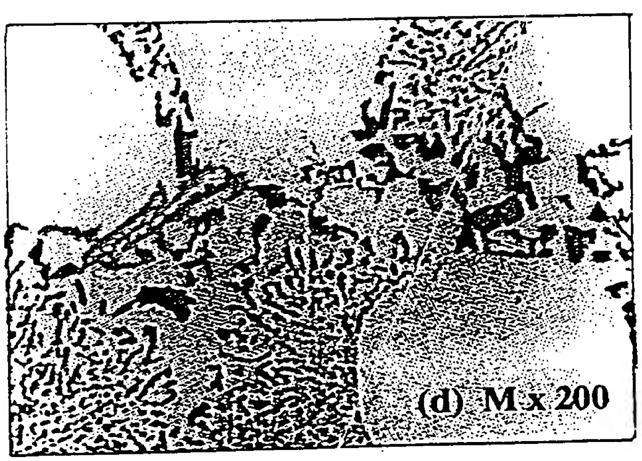


Figure 2d